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A STRUCTURAL MODEL FOR THE INTERFACE BETWEEN AMORPHOUS AND CRYST--ETC(U)
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**A STRUCTURAL MODEL FOR THE INTERFACE
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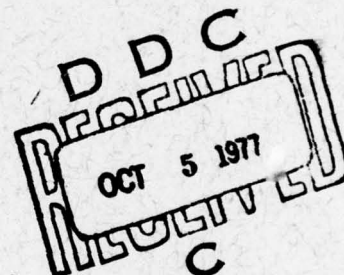
By

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Frans Spaepen

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July 1977



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Technical Report No. 2

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20. Abstract continued

radial distribution functions for each of the four layers are reported. The surface tension has a large energetic component due to the excess strain energy in both the amorphous and crystalline interface layers. For Ge the estimated surface tension is 0.235 J m^{-2} . This is used to show that the model, which contains no dangling bonds, represents a state of minimum energy. Application to the problem of creating a model for amorphous Ge by connecting randomly oriented crystallites with a random network matrix shows that such a model consists of more than 80% random network. Finally it is pointed out that the interface model is a starting point for a detailed description of the crystallization process.

* 0.235 J / sq m

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July 1977

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Division of Applied Science

Harvard University . Cambridge, Massachusetts

ABSTRACT

A general procedure for building static models of interfaces, which involves a change in phase-specific construction rules at the boundary plane, is outlined. Its application to tetrahedrally coordinated materials shows that an amorphous-crystalline interface model can be created by replacing the 'chair'-type sixfold rings (typical of the crystal) by a mixture of different ones (typical of the amorphous phase). The interface consists of two crystalline and two amorphous layers. The detailed topologies, bond angle distortions and radial distribution functions for each of the four layers are reported. The surface tension has a large energetic component due to the excess strain energy in both the amorphous and crystalline interface layers. For Ge the estimated surface tension is 0.235 J m^{-2} . This is used to show that the model, which contains no dangling bonds, represents a state of minimum energy. Application to the problem of creating a model for amorphous Ge by connecting randomly oriented crystallites with a random network matrix shows that such a model consists of more than 80% random network. Finally it is pointed out that the interface model is a starting point for a detailed description of the crystallization process.

INTRODUCTION

The crystallization of amorphous Si and Ge is a phenomenon of considerable scientific and technological interest. Recent investigations have shown that amorphous Ge crystallizes in the range 300-400°C, and the reported activation energies range from 1.4 eV - 3.5 eV (Barna, Barna and Pocza, 1972; Germain, Squelard, Bourgoin and Gheorghiu, 1975; Chik and Lim, 1975). Since specimen contamination in evaporated samples has a strong effect on the crystallization temperature and kinetics, special attention should be paid to results obtained from regrowth in amorphous Ge and Si layers produced by ion implantation of single crystals, since this technique produces very clean samples. The activation energy for crystallization in such samples is 2.0 eV for Ge (Csepregi, Kullen, Mayer and Sigmon, 1977) and 2.3 eV for Si, which crystallizes in the range 500-600°C (Csepregi, Mayer and Sigmon, 1975). All investigators agree that the magnitude of the activation energy must be explained by the formation of some intermediary defect in the amorphous-crystalline boundary during crystallization. The proposals on the nature of this defect vary from monovacancies (Csepregi, Kullen, Mayer and Sigmon, 1977), divacancies (Bourgoin and Germain, 1975), to extended divacancies (Chik and Lim, 1975). The major problem in sorting out the exact nature of this defect is the lack of knowledge about the topology of the interface structure. Therefore, in this paper a model for the amorphous-crystalline boundary in these materials will be proposed which allows detailed insight into the interface topology. As yet, no attempt will be made to describe the crystallization process

using this model, but it will provide a useful starting point to explain not only the observed activation energy but also the prefactor of the process.

GENERAL REMARKS ON THE CONSTRUCTION OF STATIC MODELS FOR INTERFACES

The interface between two phases of a one-component system is, at any given temperature and pressure, inherently unstable: the system can always lower its free energy by removing the interface and by converting to the more stable or, at equilibrium, either one of the two phases. This is obviously a problem when one wants to construct a structural model of the interface. The only completely rigorous, but cumbersome and expensive, approach to the problem is a molecular dynamics calculation on a steady state system in a small temperature or pressure gradient which stabilizes a different phase on the opposite sides of the system. (The steady state requires that a continuous flow of heat or matter is maintained.) Such a calculation would provide a complete model of planar interface in dynamic equilibrium.

An alternative approach is that of building static models by hand or with the aid of the computer. This method does not produce an equilibrium system (a static model can be considered to be at 0°K), and hence lacks the rigor of molecular dynamics. Its advantages, however, are inexpensiveness, ease of construction, and the simple physical insight it provides into the topological problem of connecting the two phases.

One way to obtain a static model of a planar interface is by first building one of the bulk phases to one side of a geometrical plane (or in other words: imagine a planar cut through the homogeneous bulk

phase, and removal of the material on one side). The construction is then continued on the other side of the plane, but a new set of construction rules, related to the structure of the bulk second phase rather than the first one, is used. If the construction rules from the second phase can be used with a plane boundary condition determined by the first phase, the structure of the subsequently added interfacial layers will gradually change, until finally the structure of the bulk second phase is reached. Such a match between the two structures should produce, if necessary after some additional local relaxation, an interface with a minimum excess energy.

It must be emphasized that this change of phase-specific construction rules at the boundary plane is an absolutely necessary feature in the construction of the static interface models. The use of more general construction principles, e.g., energy based rules, such as density maximization or distortion minimization, just leads to a continuation of the first phase beyond the boundary plane, since such general principles are common to construction of both phases. The main problem, therefore, is the choice of the appropriate phase-specific construction rules.

The most straightforward case is that involving only crystalline phases, since they are characterized completely by their lattice translation vectors. For an interface between two identical crystals of different orientation, the phase-specific rule which changes across the boundary plane is simply the orientation of the lattice vectors. It can be seen that in this case the approach outlined above leads to the construction of the symmetrical coincidence grain boundary (Gleiter and Chalmers, 1972).

When one of the phases is amorphous, the choice of the phase-specific rules is less obvious, and is in fact only possible if detailed structural information about the amorphous phase is available. The dense random packing (DRP) of hard spheres, for example, is an amorphous system, the structure of which has been thoroughly investigated (Bernal, 1964; Finney, 1970). This structural information was the basis of a recently developed static model for the interface between a DRP and a close packed crystal of hard spheres (Spaepen, 1975; Spaepen and Meyer, 1976). The construction was started with the crystalline phase, since its structure is fully known and it can conveniently be terminated on a well-defined crystal plane, e. g. (111). It was then observed that the interstices in the crystal are tetrahedra and octahedra, periodically arranged according to a ratio 2/1; the DRP, on the other hand, contains mostly tetrahedral interstices and very few octahedral ones (ratio 15/1). The occurrence of octahedral interstices was therefore taken as a phase-specific rule for the crystal, and their disappearance, together with the preponderance of tetrahedra, as a phase-specific rule for the DRP. Further construction, following the earlier described procedure and using this change of rules, resulted in the disappearance of the crystal symmetry and a gradual decrease in the localization of the subsequently added interfacial layers, until finally the DRP structure was reached.

Since this approach to modeling the crystal-amorphous interface for a hard sphere system seemed to be successful, in the sense that it explained the available direct and indirect experimental observations, an attempt will be made here to use the same basic approach to model the crystal-amorphous interface for directionally bonded, tetrahedrally coordinated systems.

SELECTION OF THE PHASE-SPECIFIC CONSTRUCTION RULES

At ambient temperature and pressure, Si and Ge crystallize into the diamond cubic structure. It is convenient to describe this structure as a stacking of layers perpendicular to a $[111]$ direction, as shown in Fig. 2c. Each atom has 4 nearest neighbors (tetrahedrally arranged, i. e., the angle between adjoining nearest neighbor bonds is exactly $109^{\circ}28'$), 12 second-nearest neighbors at 1.63 times the nearest neighbor distance (N. N. D.), and 12 third-nearest neighbors at 1.92 N. N. D., as indicated on the radial distribution function of Fig. 1. Each atom has one nearest neighbor bond parallel to the particular $[111]$ direction chosen. The three other bonds form topologically two-dimensional networks, perpendicular to this direction. These layers are not quite planar, but puckered: atoms whose fourth bond connects to the layer above are located $1/3$ N. N. D. above those whose fourth bond connects to the layer below. It is easily seen on Fig. 2c that the layers are composed of puckered six-fold rings in these "chair" configurations. In fact, the crystal can be thought of as composed entirely of these chair-type rings, since it is also the only type that is formed between successive layers. The longest distance between atoms in one particular ring is the third-nearest neighbor distance (1.92 N. N. D.). The abundance of these chair-type rings, therefore, results in a strong peak at this distance in the radial distribution function (Fig. 1).

The structure of amorphous Si and Ge has been the object of thorough investigation for the first half of this decade. By now, it has become clear that the structure can be most satisfactorily described by

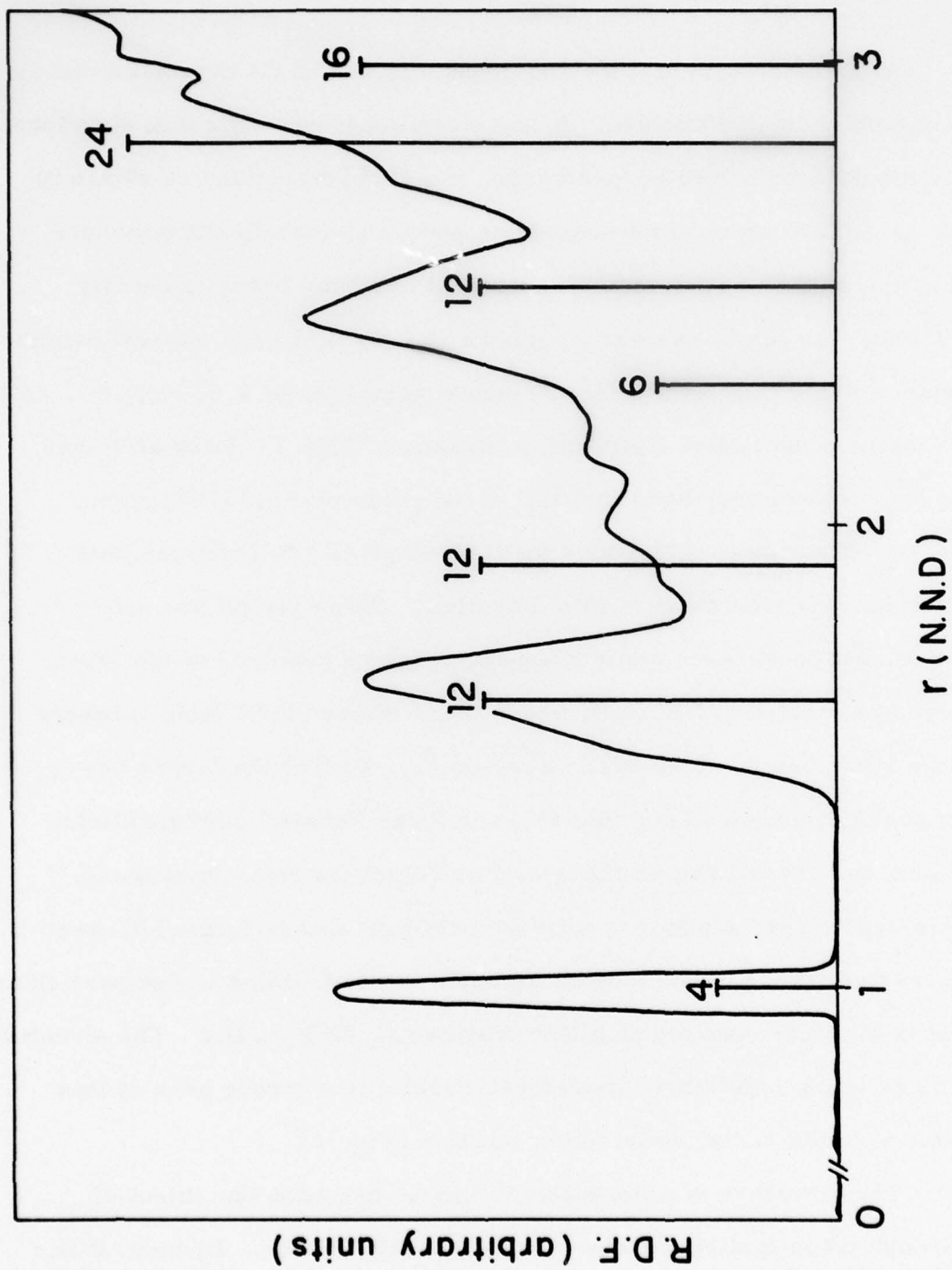


Figure 1

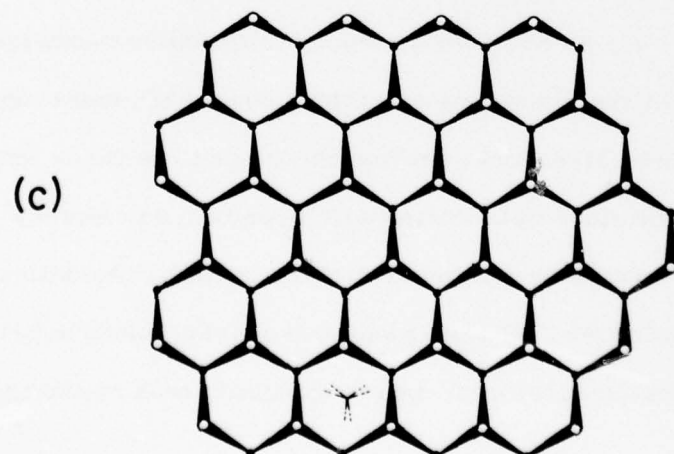
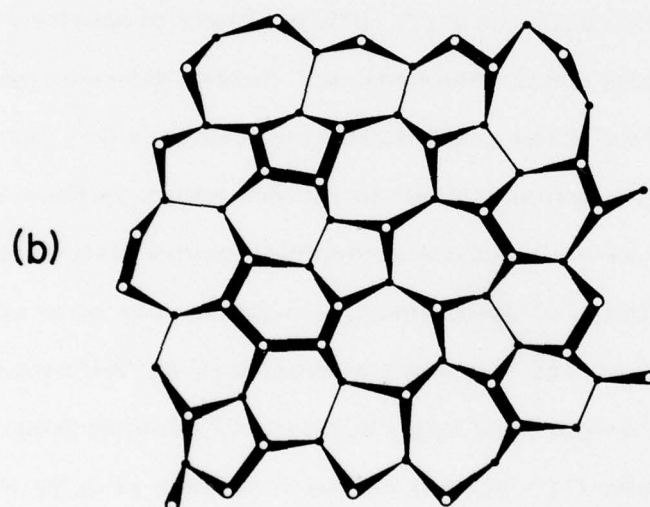
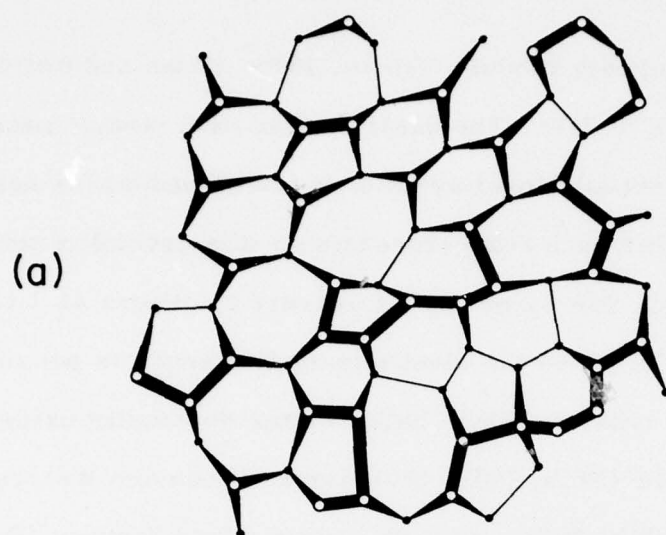


Figure 2

a continuous random network model (Moss, 1973; Moss and Adler, 1973; Paul and Connell, 1974). The basic feature of a random network is the total absence of translational symmetry, combined with a nearest neighbor environment for each atom close to that of a crystal of the same composition, i.e., for Si or Ge: 4 nearest neighbors at 1 N.N.D., and the bond angles as close to the ideal tetrahedral angle as possible. Several models of this type have been built, either physically using plastic tetrahedral units (Polk, 1971; Steinhardt, Alben and Weaire, 1974; Connell and Temkin, 1974), or with the aid of a computer (Boudreaux, Polk and Duffy, 1974; Shevchik and Paul, 1972). Their properties are in agreement with the experimental observations: density close to that of the crystal, no dangling bonds, bond angle distortion less than 20° (r.m.s. $\sim 10^\circ$), and a radial distribution function similar to the one shown in Fig. 1.

The first and second peaks of the amorphous radial distribution function are similar to those of the crystal: 4 nearest neighbors at 1 N.N.D., and 12 next-nearest neighbors at distances spread around 1.63 N.N.D. depending on the bond angle distortion. The striking difference between the two distributions is the third peak at 1.92 N.N.D., which is very strong in the crystal but almost totally smeared out in the amorphous structure. As pointed out above, this distance corresponds to the diagonal distance in the chair-type sixfold rings which make up the crystal. The amorphous structure, on the other hand, is made up of a mixture of five- to eightfold rings, in various proportions, shapes or degrees of distortion depending on the particular model. Despite these differences in ring statistics, all the random networks constructed have similar radial distribution functions, which indicates that replacing most

of the chair-type sixfold rings by any mixture of different ones will result in a disappearance of the third-nearest neighbor peak.

Based on these observations, the following phase-specific construction rules can be identified:

- (i) the chair-type sixfold ring which makes up the diamond cubic structure is characteristic of the crystalline phase;
- (ii) the virtual disappearance of these chair-type rings and their replacement by a mixture of different ones is characteristic of the amorphous phase.

CONSTRUCTION OF THE INTERFACE MODEL

The model has been constructed with the same plastic units used for bulk random networks (made by Rinco Instrument Co., Greenville, Ill.). The bonds made by these units have some flexibility which allows bond angle distortion, but are rigid enough to prevent distortions above $\sim 25^\circ$.

The description of the diamond cubic structure as a stacking of (111) layers shows clearly that the crystal can be terminated conveniently between two of these layers. Therefore, following the general procedure outlined above, it was decided to start the interface construction with the crystalline phase. A crystal consisting of two (111) layers was constructed and was terminated such that the atoms nearest to the boundary plane had one unsatisfied bond out of the plane, parallel to the [111] direction. The two layers will be designated from here on as the "1st" (closest to the boundary plane) and "2nd"

crystalline layer (see Fig. 2c). They contain 64 and 62 atoms, respectively.

At this point, the following construction rules were used in order to make the transition to the random network amorphous phase. Most important is the change in the phase-specific construction rule, which, as discussed above, consists of making the chair-type sixfold rings disappear. Specifically, this was done here by:

(i) giving preference, as much as possible within the strain limitations, to the construction of any other type of ring over a chair-type sixfold one.

Furthermore, a number of other construction rules, common to both phases were also used:

- (ii) satisfying all the bonds;
- (iii) keeping a unique nearest neighbor distance;
- (iv) minimizing the bond angle distortion.

Since bending a bond is much easier than breaking or stretching it, these last three rules form a simple energy minimization procedure. They were the criterion for choosing one type of ring over another when replacing a chair-type ring.

Since half the atoms in the first crystalline layer had at this point still one open bond, perpendicular to the plane, it was decided to continue the construction by satisfying all these bonds first. This resulted in the creation of what will be called from here on the "1st amorphous layer" (see Fig. 2b). Topologically, it is a two-dimensional structure: each of the 67 atoms, except for the layer edges of course, has three bonds with atoms in the 1st amorphous layer and a fourth bond

connecting it either downwards with the 1st crystalline layer, or upwards with the 2nd amorphous layer (to be built later).

It consists of a mixture of five-, six- and sevenfold rings. Only a few of the sixfold rings approach the "chair" configuration. Most of them have either a "boat" or a distorted configuration. It is clear from Fig. 2b that the new layer has no translational symmetry and is thus truly "amorphous". So, it seems that the particular choice of phase-specific rules made above was indeed sufficient to make the transition from crystalline to amorphous. It is interesting to note that after a few non-chair type rings have been constructed (~ 15 atoms) it was no longer necessary to pay special attention to rule (i), i. e., keep avoiding chair configurations. Once amorphousness was introduced by creating a few noncrystalline rings, adherence to the general rules (ii) - (iv) was sufficient to prevent the reappearance of crystallinity.

The second amorphous layer, formed by satisfying all the loose bonds of the first amorphous layer out of its topological plane, contains 73 atoms and is shown in Fig. 2a. Again, there was no problem keeping the structure amorphous by simply using rules (ii) - (iv). Topologically, this layer is also three-connected, except for one atom near the edge which has all four bonds in the topological plane.

Based on our previous experience with physical model building, it was felt that at this stage the construction had become qualitatively similar, in terms of the number of choices available and the amount of strain introduced, to that of a bulk random network. Therefore, it was deemed unnecessary to add any more layers. Suffice it to point out that the model can be extended in all directions without any problem. Figure 3 shows two views of the completed structure.

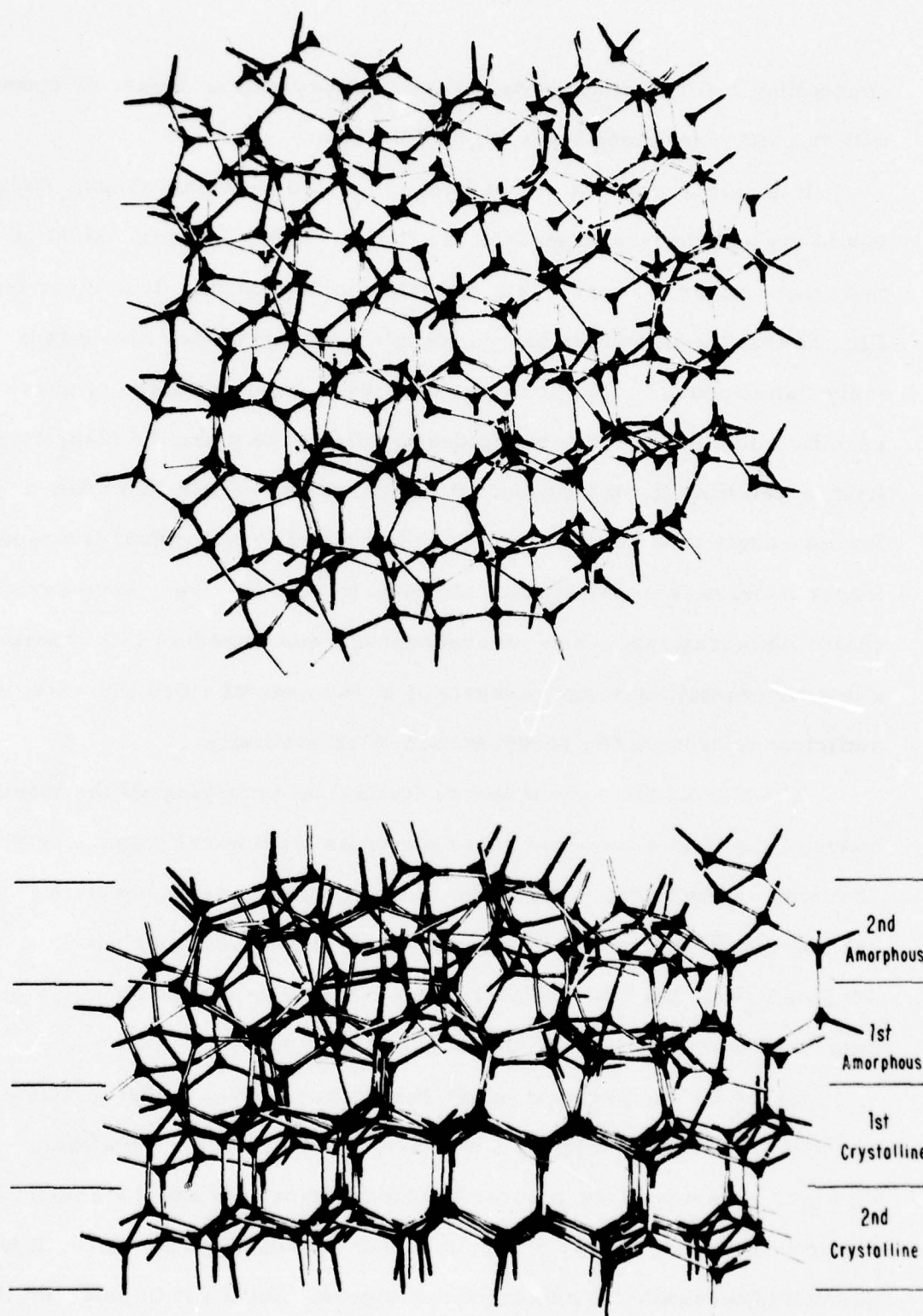


Figure 3

MEASUREMENTS AND OBSERVATIONS

The coordinates of the 266 atoms were measured by a triangulation method, using a laser mounted on a surveying transit. With the aid of the computer, using an algorithm similar to that of Polk and Boudreaux (1973), the coordinates were refined to make all the nearest neighbor bond lengths exactly equal.

The bond angle distortions could then be calculated. To illustrate the variation of the distortion with distance from the boundary plane, Fig. 4 shows the deviations ($\Delta\theta$) from the ideal tetrahedral angle of the bond angles associated with the atoms in each of the four layers parallel to the boundary. There are six angles between the four bonds of each atom; however, bond angles with an unsatisfied bond (at the model's outer surface) have obviously not been taken into account. The mean bond angle deviation $\overline{\Delta\theta}$ is close to zero in all four layers and the distributions are roughly symmetrical about the origin. The standard deviation $\sqrt{\overline{\Delta\theta^2}}$, which is a measure of the bond angle distortion, is 10.0° and 8.2° for the 1st and 2nd amorphous layer; 8.6° and 4.8° for the 1st and 2nd crystalline layer, respectively. Since this value for the bulk crystalline and amorphous phase is 0° and 9.1° , respectively, it becomes clear that the creation of the interface results in an excess distortion of the bond angles near the boundary plane. The excess strain energy associated with this distortion contributes to the surface tension. The 2nd crystalline layer is less distorted than the 1st one and it seems plausible that the excess distortion will keep decreasing with increasing distance from the boundary. The 1st amorphous layer is more

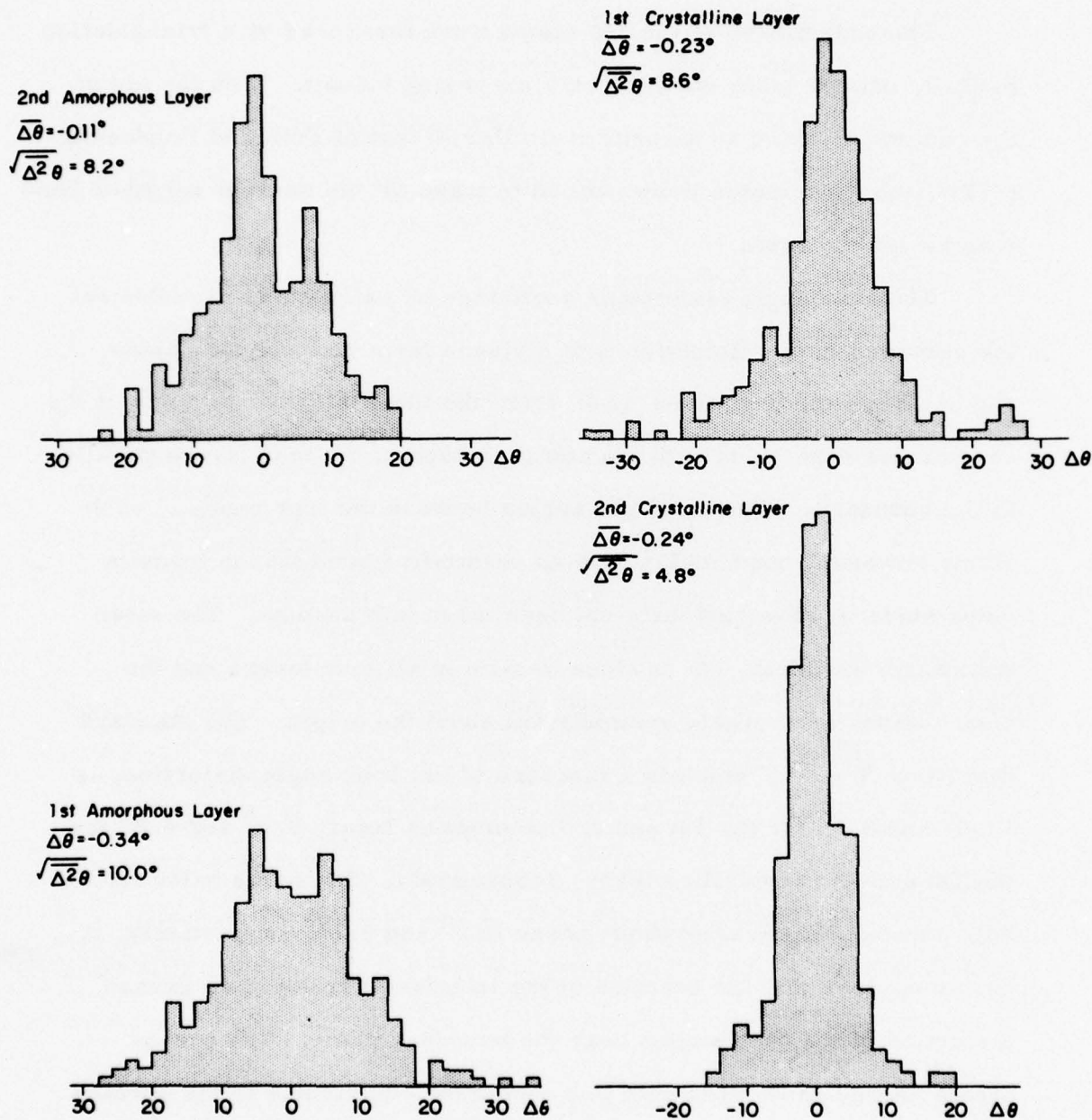


Figure 4

distorted than the bulk amorphous phase. The 2nd amorphous layer, however, seems to be less distorted; there is no obvious physical explanation for this, and it is probably an artifact caused by the large number of unsatisfied surface bonds in this layer; enlarging the model would probably result in a slight increase of the distortion. Suffice it to point out that also in this respect the 2nd amorphous layer, in agreement with the observations during the construction of the model, is quite similar to the bulk amorphous phase.

To illustrate the variation of the structure with distance from the boundary; Fig. 5 shows the radial distribution functions of the four interface layers. Since only distances between atoms in the same layer are used, these R. D. F. 's are close to two-dimensional and should be defined as $2\pi r \rho'(r)$. The average distribution is $2\pi r \rho'_0$ (i. e., linear in r), where ρ'_0 is the average number of atoms per unit area of the layer. Since the nearest neighbor distances and nearest neighbor coordination are nearly the same for all atoms in the model, their atomic volumes, defined e. g., by a Voronoi construction, are expected to be almost identical. (The difference between the bulk crystalline and amorphous atomic volume is only $\sim 1\%$). Therefore, ρ'_0 was taken to be $.866 \text{ atoms (NN D)}^{-2}$, which is the planar density of the crystalline (111) planes. Since $\rho'(r)$ is defined for an infinite layer, it was necessary to correct for the finite size of the layers by introducing a size factor $F(r, \text{size})$, which is defined as the ratio between the number of pairs at a distance between r and $r + dr$ per atom in a layer of finite size (measured on the model) and the number of pairs at the same distance per atom in an infinite layer ($= 2\pi r \rho'(r)$). Expressions for

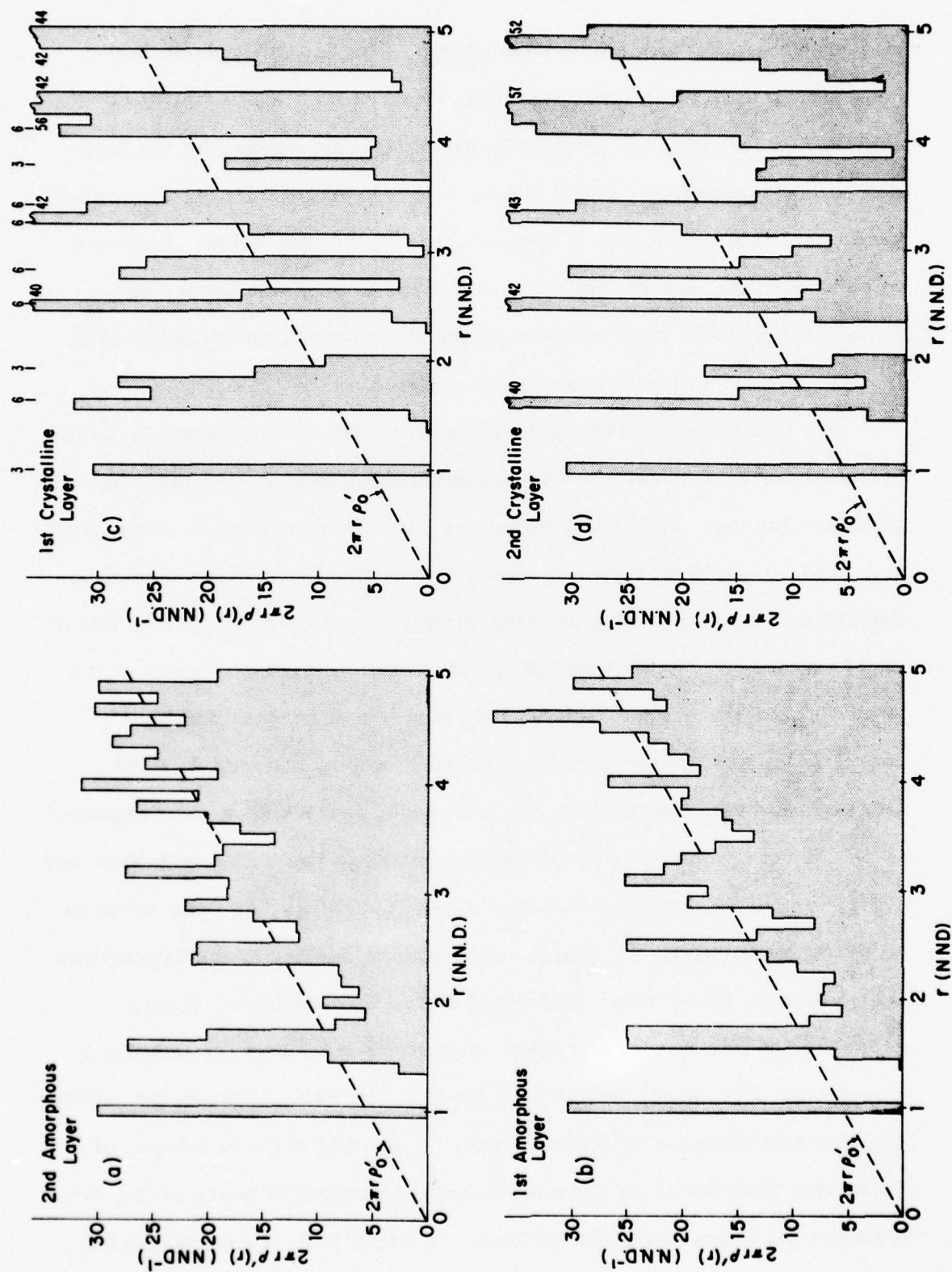


Figure 5

$F(r, \text{size})$ are available only in three dimensions, but it is easy to derive an expression in two dimensions using the same basic approach. It is assumed that the shape of the finite layers can be approximated by a circle with effective outer radius R . Formally, this means that the expression for the atom positions in a finite size layer can be obtained by multiplying the expression for an infinite cluster by a shape factor $S(r) = 1$ for $r \leq R$; $S(r) = 0$ elsewhere. It is straightforward to show that the size factor $F(r, R)$ is the autocorrelation function of the shape factor $S(r, R)$ or:

$$F(r, R) = \frac{1}{\pi} \left[2 \arctan \sqrt{\left(\frac{2R}{r}\right)^2 - 1} - \frac{r}{R} \sqrt{1 - \left(\frac{r}{2R}\right)^2} \right].$$

After this correction was made, the R. D. F. 's do indeed approach the linear average distribution at large distances and the number of nearest neighbors, obtained by integrating the first peak, is indeed very close to 3.

The R. D. F. of a perfect, undistorted crystalline (111) layer consists of a series of deltafunctions with positions and area as indicated at the top edge of Fig. 5c. The peaks in the R. D. F. 's of the 1st and 2nd crystalline layers have been broadened by the bond angle distortion, but they are in the right positions and have the right integrated peak area. The peaks of the 2nd crystalline layer are sharper than those of the 1st layer, which reflects the lower bond angle distortion. To compare the structure of the two-dimensional 2nd amorphous layer to that of the three-dimensional bulk amorphous phase, their respective pair distribution functions $w'(r) = \rho'(r)/\rho'_0$ and $w(r) = \rho(r)/\rho_0$ are

shown on Fig. 6. Because of the difference in their normalization the two functions can only be compared qualitatively, but their similarity is clear. The absence of a strong peak at 1.92 N. N. D. in $w'(r)$ reflects the phase-specific construction rule which forbids the chair-type configurations on the amorphous side of the boundary.

A final observation is concerned with the degree of localization, perpendicular to the boundary, of the amorphous layers near the interface. The layer localization can be characterized by the maximum distance normal to the boundary plane between any two of its atoms. For example, for the crystalline (111) planes, this is simply the puckering height: $1/3$ N. N. D. For the 1st and 2nd amorphous layers this was measured, after correcting for a slight overall bend in the model, to be 0.74 N. N. D. and 1.03 N. N. D., respectively. This means that the degree of localization decreases with increasing distance from the boundary, i. e.: the structure becomes more "random" in the direction normal to the boundary. The 2nd amorphous layer, however, is still more localized than a totally random distribution of the atoms would be, as can be seen from the following argument: since it is possible to establish a one-to-one correspondence between the atoms of subsequent layers, the planar density of all the layers must be the same, i. e., equal to that of a crystalline (111) plane. Since all the atoms, because of their identical nearest neighbor distance and coordination, have the same atomic volume, the spacing normal to the boundary between subsequent layers must on the average be the (111) spacing, or $4/3$ N. N. D. If the atoms in the 2nd amorphous layer were not localized at all, one would therefore expect this value to be the maximum normal distance

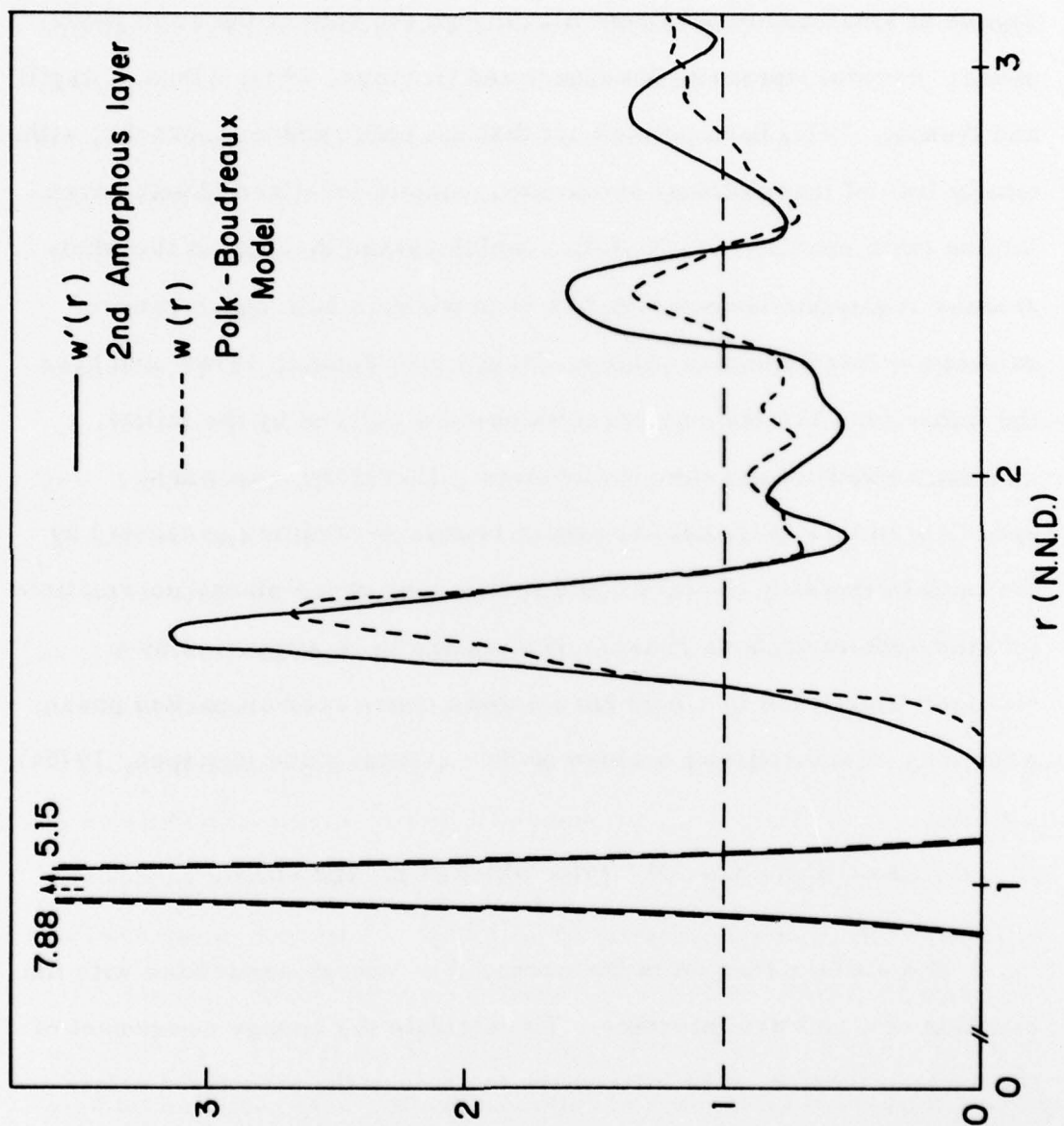


Figure 6

between any two of its atoms. Whether or not the localization of subsequent layers would disappear if the model were enlarged is open to speculation. It would not be totally surprising, however, if a certain degree of localization persisted all through the bulk of the amorphous phase. Several workers (Chaudhari and Graczyk, 1974; Alben, Cargill and Wenzel, 1976) have pointed out that the bulk random networks, although totally free of translational symmetry, contain localized planar correlations (with spacing $4/3$ N. N. D.) which extend throughout the whole model. A similar observation has been made in bulk dense random packings of hard spheres (Alben, Cargill and Wenzel, 1976), and here the indications are that the correlations are induced by the initial boundary conditions of the construction. Therefore, one might speculate in this case that the planar boundary condition presented by the initial crystalline layers could induce long range planar correlations into the bulk amorphous phase. This seems to be supported by a similar observation in a bulk hard sphere dense random packed phase, when it is interfaced with a close packed crystal plane (Spaepen, 1975a).

CALCULATION OF THE SURFACE TENSION

The surface tension is the excess free energy associated with the creation of a unit area interface. To calculate the energy component of the surface tension, it is instructive to analyze the structural origin of the energy difference between bulk crystalline and amorphous Ge.

The enthalpy of crystallization (ΔH_c) of amorphous Ge is 2.75 kcal/mole or 29.8 meV/valence electron (Chen and Turnbull, 1969).

As discussed by Polk (1971a), there are three possible contributions to the higher energy of the amorphous phase: strain energy due to bond stretching (E_s), strain energy due to bond bending (E_b), and the energy associated with the relative orientation of neighboring tetrahedra (E_d).

The first contribution (E_s) is zero for random network models with a unique nearest neighbor distance equal to the crystalline one. The second contribution has been calculated by Moss, Alben, Adler and de Neufville (1973) to be

$$E_b = \frac{3}{4} k d^2 \overline{\Delta^2 \theta} \quad \text{per electron} \quad (1)$$

where

- $\Delta^2 \theta$: the mean square bond angle deviation
 $= (9.1^\circ)^2 = (0.16)^2$ for the Polk-Boudreaux model
- d : the nearest neighbor distance
 $= 2.43 \times 10^{-10}$ m for Ge
- k : the force constant for bond bending
 $= 2.4 \text{ Nm}^{-1}$ for Ge, derived from phonon-dispersion curves.

This gives

$$E_b = 17.05 \text{ meV/electron} \quad .$$

The third contribution comes from the difference in relative rotation between neighboring tetrahedra; in the crystal, all neighboring tetrahedra are in the 'staggered' configuration, which is the one of

lowest energy; in the amorphous phase there exists a continuous distribution of all orientations between 'staggered' and 'eclipsed', which results in a higher energy. It is difficult to calculate this energy independently, and therefore it will be assumed that it simply is the balance to make the energies add up to ΔH_c , or $E_d = 12.75$ meV/electron.

The energy contribution to the surface tension can be calculated using the same approach. The results for Ge are summarized in Table 1, which shows the contribution of each of the four layers of the model. The strain energy associated with the bond angle distortion, E_b , is calculated with Equation (1). N_d is the average number of electrons per atom participating in a bond that connects two tetrahedra whose relative orientation is not the crystalline 'staggered' one. In the two amorphous layers, all four of each atom's electrons are in this case, while in the 2nd crystalline layer there are none. In the 1st crystalline layer, each atom has 3 electrons participating in bonds within the layer; the fourth electron participates in a bond with either the 2nd crystalline or 1st amorphous layer; since only the bonds to the 1st amorphous layer connect tetrahedra which are not in the 'staggered' orientation and since half the atoms in the layer are connected this way, $N_d = 0.5$ for this layer. If it is now assumed that the distribution of the relative orientations of the neighboring tetrahedra which are not in the 'staggered' configuration is the same as in the bulk amorphous phase, the resulting contribution to the energy is $E_d = N_d E_d(\text{bulk}) = N_d \times 12.75$ meV/atom. Both contributions are added up to give the total energy: $E = 4E_b + E_d$. The excess energy ΔE is the difference

TABLE 1

Layer #	$\sqrt{\Delta^2 \theta}$ (°)	E_b (meV/ electron)	N_d	E_d (meV/ atom)	E (meV/ atom)	ΔE (meV/ atom)
2nd amorphous	8.2	13.78	4	51.00	106.11	0
1st amorphous	10.0	20.67	4	51.00	133.69	14.38
1st crystalline	8.6	15.26	0.5	6.33	67.43	67.43
2nd crystalline	4.8	4.67	0	0	18.66	18.66
TOTAL						100.47

between the energy of the interfacial layers (E) and the energy of the respective bulk phases. For the crystalline layers, the bulk value is of course zero, so that $\Delta E = E$. For the amorphous layers, the bulk value is $\Delta H_c = 119.31 \text{ meV/atom}$ and $\Delta E = E - \Delta H_c$. For the 2nd amorphous layer, this results in a negative value due to a bond angle distortion (8.2°) which is apparently lower than the bulk value (9.1°). However, as discussed in the previous section, this unphysical result is probably an artifact that would disappear if another layer were added to the model. Therefore, because the 2nd amorphous layer is quite similar to the bulk amorphous phase in many other respects, its excess energy is simply set equal to zero. Adding up the contributions of the four layers gives the energy part of the surface tension: 100.47 meV/atom in the boundary plane. The area per atom in a Ge (111) plane is $6.82 \times 10^{-20} \text{ m}^2$, which means that the surface tension can be written as:

$$\sigma \geq 0.235 \text{ J m}^{-2}$$

This value for the surface tension is a lower limit. If more crystalline and amorphous layers were added to the model, they would probably also have some bond angle distortion, but their contribution to the strain energy would be small since the energy depends on the square of the distortion which decreases with increasing distance from the boundary. Another possible contribution to the surface tension is the entropy due to the localization of the amorphous layers near the boundary. However, the configurational entropy of these random networks is so low (0.2 k per atom is an upper limit (Spaepen, 1974))

that even if one complete amorphous layer would lose all its configurational entropy, the resulting change in the surface tension at room temperature would be at most 5%.

DISCUSSION AND CONCLUSIONS

1. It has been demonstrated that an interface can be constructed between the amorphous and crystalline phases of tetrahedrally coordinated materials by following the general procedure of changing the appropriate phase-specific construction rules.

2. The resulting interface has no broken bonds and this corresponds to a state of lowest energy. Breaking a bond in Ge costs 1.63 eV (Pauling, 1960), while the energy associated with forming the interface without broken bonds is .100 eV/atom in the boundary plane. This means that breaking a bond in the interface would be energetically favorable if it would relieve the strain in approximately 60 atoms around it; it was clear from the model building experience that this could never be the case. Whether or not the interfaces in crystallization experiments resemble this ideal lowest energy model depends on the extent to which loose bonds have been annealed out after preparation of the samples. It seems probable that the model would be most applicable to slow crystal growth processes, or growth in pre-annealed samples.

3. The surface tension is mainly energetic in origin. When normalized in terms of the heat of transformation per atom Δh_c , the surface tension can be written as $\sigma \geq 0.84 \Delta h_c / \text{atom}$ in the boundary

plane. This value is comparable to that calculated for the surface tension between a monatomic hard-sphere like crystal and its melt at the melting point $\sigma = 0.85 \Delta h_f / \text{atom}$ in the boundary plane. The crystal-melt surface tension, however, is negentropic in origin, which means that the surface tension scales with absolute temperature, while the amorphous-crystalline surface tension never falls below the value of its energetic component. This value is large enough to suppress homogeneous nucleation of Ge crystals in the amorphous covalent phase (Turnbull, 1969). It is difficult to check this experimentally, since most amorphous films have many impurities which can act as heterogeneous nucleation sites. The fact that other network formers (SiO_2 , Se) where one would expect a similar energetic surface tension, are good glass formers is consistent with this calculation, although in these cases it is difficult to separate the effects of jump frequency and surface tension on homogeneous nucleation.

4. The model makes it possible to be more specific about the problem of putting together an assembly of randomly oriented microcrystallites or other regular microclusters ('amorphons') by connecting them with a random network matrix (Hoare, 1976). Replacing part of a random network with a spherical crystallite of diameter d results in a lowering of the energy, since the crystallite is initially unstrained, by $\pi d^3 \Delta h_c / 6\bar{v}$ (\bar{v} is the atomic volume), and an increase in energy due to the surface tension of the newly created interface (assumed isotropic) by $\pi d^2 (0.84 \Delta h_c) / (\bar{v})^{2/3}$. The new system will have the same energy as the random network if the two contributions are equal or: $d = 5(\bar{v})^{1/3}$. Therefore, if one were to construct a microcrystalline

composite model that is energetically equivalent to the random network model for amorphous Ge, the crystallite size would be 14 \AA . The interface model shows that the 2nd amorphous layer begins to resemble the bulk phase; therefore, in order to make the transition between two randomly oriented crystallites at least three amorphous layers must be constructed between them; the middle one of those three layers approaches the bulk structure and can therefore be shared by both crystallites as their 2nd amorphous interface layer. The spacing of the interface layers is the crystalline (111) spacing d_{111} . The distance between the centers of two adjoining crystallites is therefore at least $d + 3d_{111}$. In the case of Ge, this means that the amorphous matrix necessary to join the crystallites takes up more than 80% of the volume. The only way to decrease this amount substantially is by arranging the crystallites or microclusters in a highly correlated way (Gaskell, 1975).

5. The model can be a starting point for understanding the topology and energetics of the crystallization process. It is clear that in order to make any topological changes in the model at all, it is necessary to break and reconnect the bonds. Preliminary investigations seem to indicate that it is possible to make major topological rearrangements by breaking one bond and propagating the resulting loose ends through the network; some of the intermediate configurations necessary for propagation of the loose bond require local extra strain. This strain energy could explain the difference between the energies required for bond breaking (1.63 eV for Ge and 1.82 eV for Si (Pauling, 1960)) and the activation energies of the crystallization process (2.0 eV for Ge (Csepregi, Kullen, Mayer and Sigmon, 1977) and 2.3 eV for Si (Csepregi, Mayer and Sigmon, 1975)). This question obviously needs further investigation.

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FIGURE CAPTIONS

- Fig. 1: Radial distribution function of the Polk-Boudreaux (1973) tetrahedrally coordinated random network structure. The number and positions of the neighbors in the diamond cubic lattice are indicated by the vertical lines.
- Fig. 2: Topology of the four layers in the interface model. The circles and dots represent atoms whose fourth bond connects, respectively, to the layer above or below.
- (a) the 2nd amorphous layer;
 - (b) the 1st amorphous layer;
 - (c) the 1st crystalline layer; the position of the 2nd crystalline layer is indicated.
- Fig. 3: Top and side view of the completed interface model; the four layers are indicated.
- Fig. 4: The distribution of the bond angle deviations for the four interface layers.
- Fig. 5: The radial distribution functions for the four interface layers. The dashed line corresponds to the average density. The number and positions of the interatomic distances in an undistorted (111) layer are indicated on the top edge of (c).
- Fig. 6: Comparison of the pair distribution functions of a bulk random network and the 2nd amorphous layer.

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